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ALLOY POWDER FOR POWDER METALLURGY
[Funmatsu Chikin Yo Gokin Funmatsu]

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Specification

1. Title of the invention

Alloy powder for powder metallurgy

2. Patent Claims

1. An alloy powder for powder metallurgy constituted by mixing, with the surface of an iron alloy powder comprising, in terms of wt%, not only of 0.25 ~ 0.5% of Ni & 0.25 ~ 1.0% of Mo but also, as impurities, of no more than 0.3% each of Mn & Cr and a balance of iron, a Cu and/or Mo metal powder bearing a compositional ratio of Cu: 1 ~ 3% and/or Mo: 1.0% or below with respect to the whole.

2. An alloy powder for powder metallurgy constituted by mixing, with the surface of an iron alloy powder comprising, in terms of wt%, not only of 0.25 ~ 0.5% of Ni & 0.25 ~ 1.0% of Mo but also, as impurities, of no more than 0.3% each of Mn & Cr and a balance of iron, not only a Cu and/or Mo metal powder bearing a compositional ratio of Cu: 1 ~ 3% and/or Mo: 1.0% or below but also 0.4 ~ 1.2% of a graphite powder with respect to the whole.

3. Detailed explanation of the invention

(Industrial application fields)

The present invention concerns an alloy powder for powder metallurgy, and in particular, it concerns an alloy powder for powder metallurgy ideal for manufacturing high-strength sintered low-alloy steels.

¹ Numbers in the margin indicate pagination in the foreign text.

(Prior art)

Diversified products have become available today in accordance with the advancement of the technical field of powder metallurgy, and furthermore, advanced performances have come to be required of them.

Even low-alloy sintered steels, which represent rather inexpensive materials, have come to meet unprecedented advanced performance requirements. Techniques for manufacturing low-alloy steel powders recently developed in order to meet these requirements include techniques for manufacturing Distaloy AE (Heganess Co.) & Sigmaly (Kawasaki Steel Co.). The rationale behind these techniques is the so-called “composite alloy powder” technology, whereas the main objective sought by this concept may be said to be the improvements of the compressive profile & dimensional variation rather than the improvements of strengths.

Japanese Patent Publication Kokoku No. Sho 45[1970]-9649 discloses, for the purpose of manufacturing a product which exhibits a minimal dimensional variation as well as high strengths after a thermal treatment, a method for manufacturing a low-alloy powder of the following composite profile: 0.40 ~ 1.00% Mo-0.50 ~ 2.00% Cu-1.50 ~ 2.00% Ni-Fe. /2

Japanese Patent Application Publication Kokai No. Sho 53[1978]-92306 discloses, for achieving an objective similar to the aforementioned one, a method for manufacturing a powder for powder metallurgy bearing a composite profile of 1 ~ 20% Cu-Fe. In the above, the “composite” concept represents an attempt, by heating a mixed feed powder at a high temperature within a reducing atmosphere, to partially alloy said powder by means of mutual diffusion.

It is said, on the other hand, that steel powders obtained by means of so-called “atomization” bear unsatisfactory compression profiles. In order to eradicate these shortcomings, Japanese Patent Application Publication Kokai No. Sho 59[1984]-215401 discloses & proposes a method wherein some alloy components are preliminarily mixed as atomized steel powders and wherein the remaining alloy components are mixed as a composite. The powder disclosed by the same Patent Gazette is obtained by mixing a powder bearing a profile of Ni: 2.5% or below and/or Cu: 2.0% or

below with a 0.1 ~ 1.0% Mo-Fe atomized steel powder in a state where the former is being diffused within & adhered to the atomized steel powder surface.

(Problems to be solved by the invention)

The objective of the present invention is to provide an alloy powder for powder metallurgy, above all an alloy powder for powder metallurgy suitable for manufacturing a high-strength sintered low-alloy steel bearing an excellent tenacity.

(Mechanism for solving the problems)

Incidentally, the present inventors compiled various investigations for the purpose of achieving the above-mentioned objective, as a result of which the following aspects were discovered to be essential for obtaining so-called "high-strength sintered low-alloy steels," including thermally treated versions of sintered materials:

①: The collective strength of a sintered product is determined by the strength of a foundation and the strength of a sintered junction unit, and it is important to balance both. It is necessary, due to the structural peculiarity of the sintered product, for the material strength of the sintered junction unit to be higher than the foundation strength. Incidentally, the junction unit strength also depends on the junction unit shape.

In other words, the combination of a feed powder, namely a base steel powder determining the foundation strength, and an additive powder represents one of the critical factors determining the product strength.

②: It is desirable for the tissue of the sintered junction unit to be homogeneous.

In other words, it is important to select combinations of base steel powders & additive metal powders in consideration also of the effects of mixing states, diffusion phenomena, etc.

Figure 1 through Figure 3 are abstract demonstrational diagrams pertaining to the sintered portion of a product obtained by sintering a metal powder.

As Figure 1 indicates, the sintered product of the metal powder is constituted by the foundation unit (1) prevailing as a base comprising of the feed powder and the sintered junction unit (2) formed as a result of sintering. Sites painted plainly in black represent pores. It is thus apparent that the mechanical properties of the sintered product are determined by the strength of the foundation inclusive, as a base, of the feed powder (hereafter referred to as the "foundation strength") and the strength of the sintered junction unit (hereafter referred to as the "junction unit strength").

The foundation strength is determined mainly & almost exclusively by the composition of the feed powder, namely base steel powder, and by materials formed by diffused elements.

The constituent materials of the sintered junction unit, on the other hand, are determined by elements diffused and/or melted within the junction unit, and the strength of the sintered junction unit is affected by these constituent materials of the junction unit.

The sintered junction unit, furthermore, generally forms an extremely thin bridge in comparison with the foundation unit, as Figure 1 indicates, and since the sintered junction unit strength is affected by this shape, a destruction outbreak route for the sintered product tends to be formed.

From the standpoint of improving the mechanical performances of a sintered product, therefore, it is important to fortify, in consideration also of the junction unit shape, the sintered junction unit strength and to select a foundation strength well-balanced with the junction unit strength. In this case, not only does the sintered junction unit form, as has been discussed earlier, a fine bridge but, as the notation A indicates in the figure, depressions on particle joining lines also tend to persist, and since porous portions virtually serve as internal defects & notches, a destruction outbreak route tends to be formed. In terms of material-specific strengths, too, therefore, it is desirable for the strength of this sintered junction unit to be higher than that of the foundation unit. As far as the shape & pore size of the sintered junction unit are concerned, furthermore, a thicker

bridge and a smaller & rounder pore are presumed to be respectively desirable for the former and latter.

In other words, it is necessary, at least in an ideal setting, for the sintered junction unit (2), preferably the peripherals of the pores (3) thereof, to be homogeneously alloyed and strengthened, as Figure 2 indicates. In this case, it is necessary not only for the tissue of the sintered junction unit (2) to be homogeneously alloyed and strengthened but for the strength of the foundation unit (1) /3 to be also designated at a corresponding level in a well-balanced manner.

As Figure 3 indicates, on the other hand, even if attempts are made to up the strength of the sintered junction unit (2) by adding an alloy component(s) in a case where the sintered junction unit (2) is divided into a zone (2') strengthened more than necessary & a zone (2'') not sufficiently strengthened, destruction progresses along sites with low strengths, as the bold broken line in the figure indicates, due to the inevitability of the proliferation of destruction through weak sites.

Incidentally, the sintered product targeted by the present invention is processed in massive quantity on an industrial basis, and for this reason, it must satisfy, apart from the above-mentioned mechanical performance requirements, the following requisites as well:

(i): That it be usable within sintering & thermal treatment atmospheres being generally orchestrated on an industrial basis, namely that it be specifically usable even within an RX atmosphere inclusive, as an atmospheric component, of some oxidizing gases, and that it entail sufficient manifestations of strengths even at a sintering temperature of 1,130°C, above all that it yield excellent performances under conditions prevailing after the molding, sintering, & thermal treatments.

(ii): A component-invariable system bearing a higher compressive profile & a lower porosity can be strengthened more easily under identical molding conditions, and thus, a superior compressive profile is preferred.

(iii): That dimensional variations of a sintered product and a sintered & thermally treated product be minimal, and this value must be below the $\pm 0.4\%$ threshold with regard at least to practical applications.

Various investigations & experiments were repeated for the purpose of achieving these objectives, as a result of which the strengthening goal has been realized by adding Ni & Mo to a low-alloy steel inclusive of Mn & Cr, whereas, on the other hand, the goal of forming a homogeneous tissue has been realized by pre-alloying Ni. It was discerned, on the other hand, that, in a case where other metal components such as Cu, Mo, etc. are added as a composite, goals of T.S. ≥ 95 kgf/mm² & Ch ≥ 1.1 kgf/mm² can be realized even if an RX gas low-temperature sintering [or] RX gas thermal treatment ($1,130^{\circ}\text{C} \times 20$ min.; CO₂ $\geq 0.3\%$) is performed, based on which the present invention has been completed.

Thus, the marrow of the present invention lies in an alloy powder for powder metallurgy constituted by mixing, with the surface of an iron alloy powder comprising, in terms of wt%, not only of 0.25 ~ 0.5% of Ni & 0.25 ~ 1.0% of Mo but also, as impurities, of no more than 0.3% each of Mn & Cr and a balance of iron, not only a Cu and/or Mo metal powder bearing a compositional ratio of Cu: 1 ~ 3% and/or Mo: 1.0% or below but also 0.4 ~ 1.2% of a graphite powder with respect to the whole.

It is desirable, in a preferred embodiment, for the particle size of the base steel powder to be 250 μm or less, and as for the additive alloy powders, it is desirable for the sizes of the Cu & Ni powders to be each 44 μm or less and for the size of the Mo powder to be 10 μm or less (diameter).

(Functions)

The above-mentioned constitution is borne by the alloy powder of the present invention, and in the following, reasons for thus limiting the powder composition will be explained in further detail.

Since sintering & thermal treatments within an RX atmosphere must be possible as a requisite mandated by the objective of the present invention, a composition comprising mainly of Ni, Cu, Mo is used in the present invention.

The base steel powder bears profiles of Ni: 0.25 ~ 0.5% & Mo: 0.25 ~ 1.0%, whereas effects of improving tenacity & quenchability can be achieved in cases where the respective ratios of Ni & Mo are confined to the aforementioned ranges.

Reasons behind the addition of Ni in a pre-alloyed state lie not only in an effective improvement of the foundation strength but, as is shown below, also in the problem of the addition of metallic Ni in a composite state, which has been newly discovered. In other words, in a case where an Ni-Cu component is added in a composite morphology, especially in the course of an Rx gas sintering treatment at 1,130 ~ 1,150°C, not only mixing heterogeneity but also tissue heterogeneity attributed to diffusions among the respective elements of Ni, Cu, & Fe arise, as a result of which a strength loss attributed to the strength heterogeneity of the sintered junction unit shown earlier in Figure 3 becomes unavoidable. A strength loss attributed to the generation of pores accompanying a phenomenon estimated as a manifestation of the Kirkendall effect also becomes incurred.

The additions of Mo based on pre-alloyed & composite morphologies are both effective, although different strength contributory mechanisms are involved in the respective methods. In other words, the addition of Mo in a pre-alloyed state is effective for improving the foundation strength. It is presumed to contribute to the strengthening of the sintered junction unit based on diffusion.

Mn & Cr:

0.3% or below each of Mn & Cr are permissible as impurities, whereas in a case where this threshold is exceeded, oxidation becomes induced in the course of sintering even within an RX gas atmosphere, as a result of which performance deteriorations of the product become unavoidable. /4

As far as metal components added in a composite morphology are concerned, 1.0 ~ 3.0% of Cu may be internalized. Cu exerts effects not only of accelerating the generation of a sintered junction unit based on the generation of a solid phase but, as an intrinsic attribute of Cu, also of improving the sinterability of the Fe-C system.

Moreover, it becomes possible, by adding 0.2 ~ 1.0% of Mo as a composite additive metal component, to improve the hardness & strength. The addition of Mo abiding in a composite morphology is effective for improving the hardness & strength of the junction unit, although it is desirable for the particle size & internalization ratio to be designated respectively at 10 µm or below & 1.0% or below, for the intrinsic solid solubilization tendency of Mo is minimal.

It is desirable to add Mo & Cu as a mixture. The reason is because the addition of Mo is especially effective in a case where a liquid phase of Cu is generated.

The effect of the carbon internalization ratio onto the strength is rather minor, whereas the alloy component system of the present invention is peculiarly characterized by the fact that a high hardness & a high strength can be maintained within a high carbon region. It therefore becomes possible, in a case where the graphite addition ratio is confined to a range of 0.4 ~ 1.2%, to induce the manifestations of the aforementioned characteristics.

In a case where the aforementioned iron alloy powder is mixed with the metal powder inclusive of Cu and/or Mo, they may be simply mixed, or such a mixture may instead be further heated, partially diffused, & alloyed within a non-oxidative atmosphere (e.g., hydrogen atmosphere, etc.) for inducing their adhesion to the steel powder surface, or the same may instead be adhered by using an appropriate binder.

The feed powder thus obtained is then molded, sintered, and, if necessary, thermally treated, whereas routinely established conditions suffice for these treatments. According to the present invention, in particular, a compressed powder can be sintered as a result of retention within an RX gas atmosphere at 1,130 ~ 1,150°C for 20 ~ 30 min., based on which it becomes possible to sinter the same at a relative low temperature & a high dew point [sic]. The atmosphere of such a case may

include approximately 0.3 ~ 0.5 vol% of gaseous CO₂, which represents a type of oxidizable gas. The sintered product thus obtained is retained within the RX gas atmosphere at 800 ~ 950°C for 60 min., rapidly cooled within an oil, and, if necessary, annealed within an open atmosphere.

Next, the present invention will be explained more concretely with reference to application examples.

Application Example 1

A steel powder bearing low oxygen & low carbon contents was prepared based on the oil atomization-decarbonizing method according to each of the compositions shown in Table I, and it was then subjected to molding, sintering, and then thermal treatments under the following conditions:

- (1): Mixing & molding conditions: Preparation & molding of a 0.8% Zn-St-added, 0.6% C-mixed, & 2% Cu-mixed compressed powder bearing a density of $7.10 \pm 0.05 \text{ kg/cm}^3$;
- (2): Sintering conditions: Sintering within an RX gas (CO₂ = 0.4%) at 1,130°C for 20 min.;
- (3): Thermal treatment conditions: Thermal treatment within an RX gas (carbon potential: 0.6%) at 850°C for 60 min. and then oil quenching; heated within an open atmosphere at 180°C for 90 min., followed by spontaneous cooling.

The powder profiles, compressed powder profiles, & sintered product profiles were evaluated on the respective occasions, and the obtained results are shown & summarized in the same table.

Incidentally, the average particle size of each base steel powder was 75 µm (diameter).

Table I

/5

No.	化 学 成 分 (wt%)						粉 体 的 性		压 粉 特 性 *			Sintered Product Properties			備 考
	Mn	Cr	Ni	No	C	F	R	A.D. (μ/cm^2)	F.R. (sec/°)	P (kg/cm ²)	R _s (%)	T.S. (kgf/cm ²)	G _t (kgf/cm ²)	吸水率 (%)	
1	0.15	—	0.25	0.26	0.002	0.025	0.0001	3.14	20.6	7.09	0.48	92.7	1.35	0.30	本 発 明 の 例
2	0.15	—	0.25	0.58	0.003	0.024	0.0001	3.03	22.3	7.03	0.67	101.4	1.45	0.26	
3	0.13	—	0.26	1.00	0.001	0.018	0.0001	3.11	21.5	7.00	0.79	100.4	1.38	0.27	
4	0.14	—	0.48	0.25	0.001	0.016	0.0001	3.08	21.8	7.07	0.68	98.6	1.37	0.28	
5	0.15	—	0.50	0.66	0.002	0.030	0.0003	3.12	21.5	7.05	0.80	104.8	1.48	0.25	
6	0.15	—	0.51	0.99	0.001	0.017	0.0001	3.11	21.3	7.01	0.71	103.4	1.50	0.27	
7	0.14	—	—	—	0.003	0.025	0.0002	3.02	22.0	7.11	0.69	78.4	0.71	0.30	
8	0.15	—	—	0.26	0.001	0.016	0.0003	3.09	21.7	7.10	0.54	86.8	1.12	0.32	
9	0.16	—	—	0.52	0.001	0.027	0.0003	3.13	20.9	7.09	0.56	89.8	1.34	0.34	
10	0.18	—	—	1.00	0.002	0.018	0.0002	3.11	21.5	7.07	0.61	93.1	1.37	0.29	高 温 で 強 度 が 保 持 さ ず と く な い
11	0.14	—	—	1.48	0.001	0.024	0.0001	3.14	21.0	6.96	0.79	96.2	1.39	0.31	
12	0.15	—	0.25	—	0.001	0.025	0.0002	3.12	21.0	7.10	0.52	86.2	1.09	0.28	
13	0.15	—	0.52	—	0.003	0.021	0.0002	3.12	21.0	7.09	0.70	87.1	1.18	0.33	
14	0.16	—	0.74	0.24	0.001	0.016	0.0001	3.06	22.0	7.03	0.78	88.8	1.15	0.27	
15	0.13	—	0.76	0.52	0.001	0.017	0.0003	3.08	21.8	7.02	0.60	92.3	1.21	0.28	
16	0.15	—	0.75	0.98	0.002	0.022	0.0003	3.15	20.9	6.95	0.78	93.5	1.23	0.29	
17	0.14	—	1.03	—	0.002	0.016	0.0001	3.11	21.3	7.04	0.72	88.4	1.08	0.31	
18	0.15	—	1.02	0.55	0.003	0.018	0.0001	3.06	22.2	6.97	0.82	87.7	1.04	0.33	
19	0.15	—	1.00	0.72	0.001	0.024	0.0001	3.13	20.8	6.94	0.90	83.2	1.07	0.28	
20	0.16	—	1.54	—	0.001	0.022	0.0003	3.15	20.9	6.95	0.84	83.4	0.98	0.29	

[(1): Chemical components; (2): Powder properties; (3): Compressed powder properties*; (4): Sintered product properties; (5): Dimensional variation; (6): Miscellany; (7): Example of the present invention; (8): Comparative example; (9): Effects attributed solely to pre-alloying of Mo minimal; (10): Compressive profile loss due to high addition ratio; (11): Attenuated effects due to excessively high pre-alloyed Ni contents; (12): Oxidation-free upper limits for Mn & Cr; (13): Performance losses due to oxidations of Mn & Cr; [Notes]: *: 0.8 wt% Zn-St added, 5 ton/cm² mold, 11.3H x 11.3H]

The respective performances shown in Table I are summarized in the triangular diagrams /6 of Figures 4 ~ 6.

The compressed powder density drops significantly in a case where the pre-alloyed alloy element contents exceed 1 ~ 1.5%. The segment surrounded by bold lines in Figure 4 represents favorable results.

An enlarged Ni content beyond 0.5% is not very effective for improving the tensile strength. The segment surrounded by bold lines in Figure 5 represents a high-strength region of 95 kgf/mm² or higher.

An excessive pre-alloying of Ni exerts adverse effects on the impact value, as in the case of T. S. The segment surrounded by bold lines in Figure 6 represents a high-strength region of 1.10 kgf/mm² or higher.

The common regions surrounded by bold lines in Figures 4 through 6 represent high-performance regions bearing compressed powder densities of 7.00 g/cm³ or higher, tensile strengths of 95 kgf/mm² or higher, and impact values of 1.10 kgf/mm² or higher. In other words, 0.25 ~ 0.50% of Ni & 0.25 ~ 1.00% of Mo are desirable.

In a case where the Mn Cr internalization ratios are each 0.30% or lower and where their sum is 0.60% or lower, as in the cases of sample Nos. 21 ~ 28, favorable performances are exhibited within the RX gas atmosphere without entailing oxidation, although their higher internalization ratios entail oxidation in the course of sintering, and significant sintered product performance losses are acknowledged.

Application Example 2

The functions & effects of metal powders mixed with the base steel powder were evaluated in the present example.

The prescribed metal powder compositions are shown & summarized in Table II. The base steel powder composition was 0.5% Ni-0.5% Mo-Fe. Cu, Ni, & M were each added as uncompounded metal powders. The compositions were provided by means of simple mixing.

Table II

No.	系	添加粉(%)		
		Cu	Ni	Mo
1	—	0	0	0
2	Cu	0.5	—	—
3		1.0	—	—
4		2.0	—	—
5		3.0	—	—
6		4.0	—	—
7		1.0	2.0	—
8		2.0	2.0	—
9		1.0	4.0	—
10	Ni	2.0	4.0	—
11		1.0	—	0.5
12		2.0	—	0.5
13	Mo	3.0	—	0.5
14		4.0	—	0.5
15	Cu	1.0	2.0	0.5
16		1.0	4.0	0.5
17	Mo	2.0	2.0	0.5

注)・組成金体に対する割合の
ベース鋼粉平均粒径15μm
(直径)

No.	系	添加粉(%)		
		Cu	Ni	Mo
18	Cu	2.0	2.0	0.6
19	Mo	2.0	4.0	0.5
20	—	—	1.0	—
21	Ni	—	2.0	—
22	—	—	4.0	—
23	Ni	—	1.0	0.5
24	—	—	2.0	0.5
25	Mo	—	4.0	0.5
26	—	—	—	0.5
27	Mo	—	—	1.0
28	—	—	—	1.5
29	Cu	2.0	—	1.0
30	—	2.0	—	1.5
31	Mo	2.0	—	2.0

[(A): System; (B): Additive powder (%); [Note]: Ratio (%) with respect to the entire composition;
base steel powder average particle size = 75 μm (diameter)]

Each of the feed materials thus prescribed was molded, sintered, and then thermally treated according to procedures comparable to those in Application Example 1.

The mechanical performances of the sintered & thermally treated products thus obtained were tested & evaluated. On an occasion for evaluating the obtained results, criteria of T. S. \geq 95 kgf/mm² and Ch \geq 1.1 kgf/mm² were established for selecting optimal additive component systems, as a result of which it was determined that the aforementioned criteria can be met only in a case where a Cu and/or Mo metal powder bearing a compositional ratio of Cu: 1 ~ 3% and/or Mo: 1.0%

or below is added as a composite metal powder, as the representative examples summarized in Figures 7 through 9 indicate.

As Figure 9 indicates, it is impossible to meet the aforementioned criteria in a case where Ni is mixed as a composite metal powder.

The favorable results of the case where the Cu powder is added and the insufficient strengths of a more advanced alloy system, namely Ni-Cu, are attributed presumably to the heterogeneity of the sintered junction unit, as has been discussed earlier.

A sintered & thermally treated product wherein 3.5% of an Ni powder and 2% of a Cu powder had been added to the base steel powder, for example, yielded a sintered junction unit comprising of a heterogeneous tissue. Along the powder interface, in particular, a segment characterized by a high Ni-Cu concentration exists, and the diffusions of these alloy elements /7 were judged to be insufficient. Cu, on the other hand, was diffused in a virtually homogeneous fashion within the tissue of a sintered & thermally treated material wherein 2% of a Cu powder had been added to the base steel powder, and the pervasion of a homogeneous annealed martensite tissue was acknowledged.

In a case where Ni is added, together with Cu & Mo, in a composite morphology, the respective mixing homogeneities of Ni, Cu, Mo, & Fe particles may be presumed problematic, although the attenuation of the diffusion power of Cu into Fe particles intrinsic to the component system is also conceivable within the above-mentioned low-temperature sintering region.

In other words, the diffusion of Cu within Ni is faster than the diffusion of the same within Fe, as a result of which its diffusion within iron powder particles becomes insufficient, and a heterogeneous tissue corresponding to the existence of Ni presumably arises. The mutual diffusion of Ni into Cu, furthermore, is fast, and the solid solution limit is also high. The melting point of the Cu-Ni system becomes elevated as a result of the solid solution of Cu & Ni.

Thus, the above-mentioned adverse effects of the Cu-Ni system combination on the diffusion potential of Cu within Fe particles become triggered in a case where the sintering

temperature is low, and since the diffusion is insufficient, it is difficult to form a homogeneous tissue, which is in turn presumed to entail strength losses.

Figures 10 and 11 respectively show tissue micrograms of a sintered product obtained by adding 2% of a Cu powder to a 0.5% Ni-0.5% Mo-Fe steel powder and a sintered product obtained by adding 3.5% of an Ni powder, 2% of a Cu powder, & 0.5% of an Mo powder to a 0.5% Ni-0.5% Mo-Fe steel powder.

The magnifying ratio is $\times 100$ in Figure 10 (a) & Figure 10 (b), whereas the magnifying ratio is $\times 500$ in (b) of the same figures.

Figures 10 pertain to the present invention, from which the pervasion of a homogeneous tissue can be inferred. The pervasion of a heterogeneous tissue wherein pores are elongated in a crevasse form attributed to the pervasion of the Ni powder in a composite morphology can be inferred from Figures 11.

Application Example 3

Effects of the graphite addition ratio on mechanical performances were examined in the present example. Mixing, molding, sintering, & thermal treatment conditions were comparable to those in Application Example 1.

The respective variations of the tensile strength, Charpy impact value, and hardness of a case where the graphite internalization ratio of the mixed powder corresponding to No. 5 in Table I was variously designated were investigated.

The results are shown & summarized in the respective graphs of Figures 12, 13, & 14.

For comparative purposes, furthermore, the performance variations of a case where the graphite addition ratio of a commercial powder bearing the composition shown in Table III was likewise designated variously were investigated. The results are likewise shown & summarized in Figures 12, 13, & 14.

It can be seen that strength, impact value, & hardness losses can each be avoided within a high carbon region of 0.4 ~ 1.2% according to the present invention. A high-strength sintered material can be obtained by adding graphite.

Table III

	C	O	Mn	P	S
市販鋼粉	0.011	0.081	0.04	0.005	0.006

Cr	Ni	Mo	Co	圧粉体特性	
				σ (g/cm ²)	g/cc
< 0.01	3.80	0.40	1.37	7.01	0.71

[(A): Commercial steel powder; (B): Compressed powder performances]

(Effects of the invention)

As the foregoing explanations have shown in detail, it becomes possible, according to the present invention, to obtain high-performance low-alloy sintered steels bearing compressed powder densities of 7.00 g/cm³ or higher, tensile strengths of 95 kgf/mm² or higher, & impact values of 1.10 kgf/mm² or higher even within an RX atmosphere inclusive, as an atmospheric component, of some oxidizable gas such as CO₂ ≥ 0.3% and even in a case where a sintering treatment is performed at a low temperature over a brief period, as in the case of sintering temperature = 1,130°C × 20 min., and thus, the present invention promises immense practical effects & commercial benefits.

4. Brief explanation of the figures

Figure 1 through Figure 3 are abstract demonstrational diagrams pertaining to the sintered segments of products obtained by sintering metal powders;

Figure 4 through Figure 6 are triangular diagrams summarizing the compressed powder densities, tensile strengths, & impact values of an application example of the present invention;

Figure 7 through Figure 9 are graphs showing the effects of the respective composite morphologies of Cu, Mo, & Ni;

Figure 10 (a), Figure 10 (b), Figure 11 (a), & Figure 11 (b) are metal tissue micrograms of sintered products; and

Figure 12, Figure 13, & Figure 14 are each graphs likewise summarizing the data of an /8 application example of the present invention.

(1): Foundation unit; (2): Sintered junction unit; (3): Pores.

Applicant: Sumitomo Metal Industries Co., Ltd.

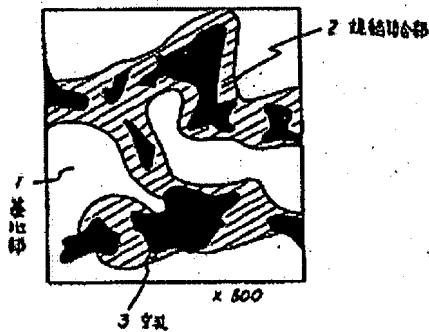
Agent: Shoichi Hirose, patent attorney

Figure 1



[(1): Foundation unit; (2): Sintered junction unit; (3): Pores]

Figure 2



[(1): Foundation unit; (2): Sintered junction unit; (3): Pores]

Figure 3



[(A): Destruction; (2'): High-strength junction unit; (2''): Low-strength junction unit]

Figure 4

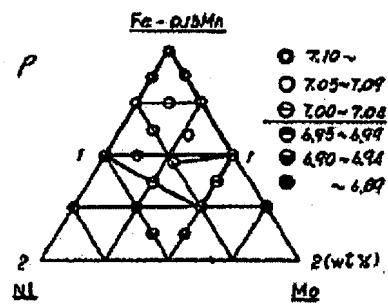


Figure 5

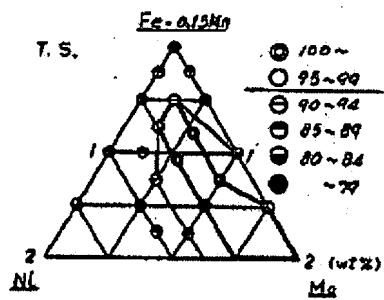


Figure 6

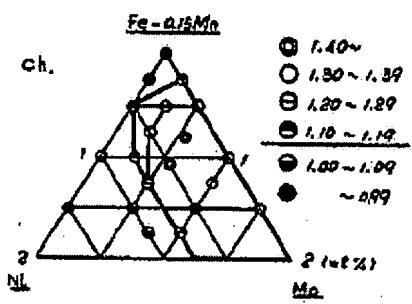
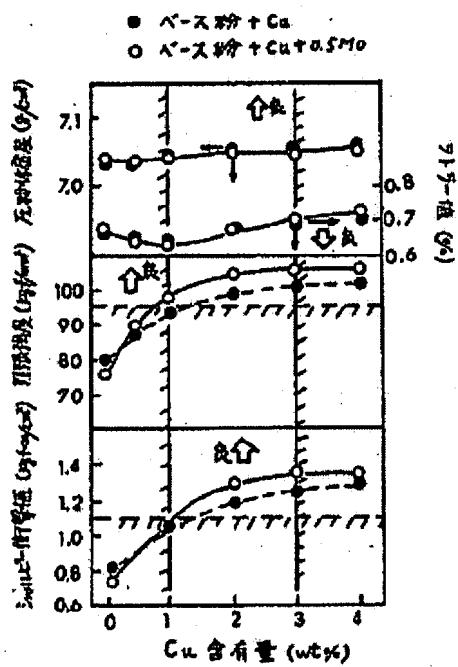


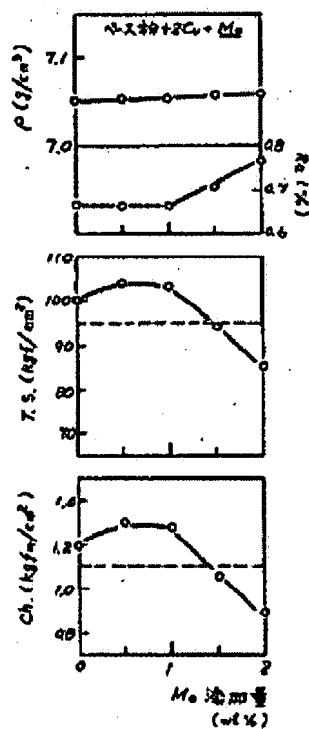
Figure 7

/9



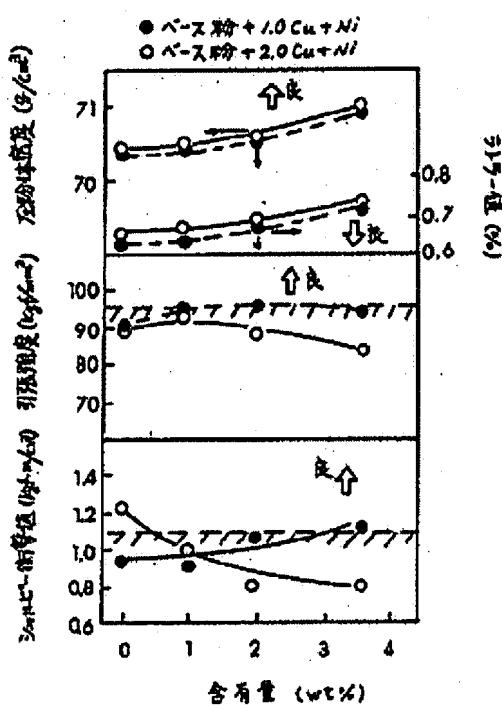
[●: Base powder + Cu; ○ base powder + Cu + 0.5Mo; (A): Compressed powder density; (B): Tensile strength; (C): Charpy impact value; (D): Rattler value; (E): Favorable; (F): Cu internalization ratio]

Figure 8



[(A): Base powder + 2Cu + Mo; (B): Mo addition ratio]

Figure 9



[●: Base powder + 1.0 Cu + Ni; ○: Base powder + 2.0 Cu + Ni; (A): Compressed powder density; (B): Tensile strength; (C): Charpy impact value; (D): Rattler value; (E): Favorable; (F): Internalization ratio]

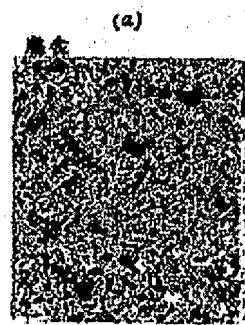
Figures 10



[(A): Corrosion]

Figures 11

(a)



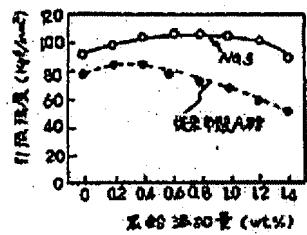
$(0.5Ni - 0.5Mo) + 3.5Ni + 2.0Cu$
+ 6.5 Mo

(b)

[(A): Corrosion]

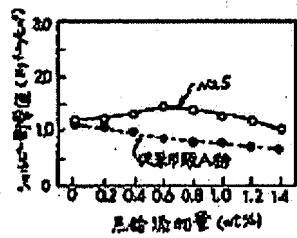
Figure 12

/10



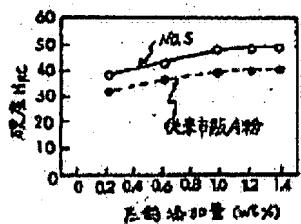
[(A): Tensile strength; (B): Graphite addition ratio; (C): Commercial powder A of the prior art]

Figure 13



[(A): Charpy impact value; (B): Graphite addition ratio; (C): Commercial powder A of the prior art]

Figure 14



[(A): Hardness; (B): Graphite addition ratio; (C): Commercial powder A of the prior art]